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Strain hardening in $\text{Ni}_3(\text{Al}, \text{Hf})$

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Abstract

A method of repeated stress relaxations is used to determine the strain-hardening that takes place during stress relaxation tests performed at different temperatures and for various crystallographic deformation axes. It is shown that strain hardening for stress relaxation conditions is substantially higher than for constant strain-rate deformation conditions. This behaviour is found to be more and more pronounced with increasing temperatures. It is suggested that strain-hardening is not dislocation substructure dependent but is inherent to dislocation motion.

1. Introduction

Several intermetallic compounds with the L_{12} structure such as $\text{Ni}_3(\text{Al}, \text{Hf})$ exhibit a positive temperature dependence of the yield stress up to a peak temperature (T_p) above which it finally decreases (for reviews see [1, 2]). This strength improvement with increasing temperatures is of particular interest for industrial applications, especially in aeronautics [3]. For instance, the nickel-base superalloys used for turbine engine components are strengthened by L_{12} precipitates. Therefore, this anomalous behaviour is the subject of extensive experimental works and theoretical modelings that attempt to describe the controlling mechanism of plastic flow, both below and above the peak temperature. The models that have been proposed to explain the flow stress anomaly generally consider a cross-slip of the screw superdislocations from $\{111\}$ planes, where they are glissile, onto $\{001\}$ planes where they become sessile. In these models, the cross-slip process is thermally activated and therefore responsible for the observed increase of the yield stress [4 - 6].

It has also been suggested that strain hardening can play an important role in determining the flow stress in Ni_3Al [2, 7, 8]. Studies of the strain-hardening rate ($\theta = d\tau_a/d\gamma$, where τ_a denotes the applied stress) in L_{12} compounds have shown that θ exhibits as well an anomalous behaviour with temperature. It has been reported that θ exhibits one or two peaks with temperature depending on the deformation axis [9, 10]. Cottrell-Stokes type temperature change experiments have been conducted in an attempt to characterise the thermal reversibility of the flow stress in Ni_3Al . To date, the interpretation of these last experiments is controversial [2, 11].

It is well known that a stress relaxation test yields an apparent activation volume (V_a) which must be corrected by a strain hardening correction term (V_h) in order to obtain the true microscopic or "effective" activation volume (V_{eff}) [12]. Recently, a method of successive stress relaxations has been developed to measure V_{eff} [13]. This technique has been successfully used to characterise a critical stress for macroplastic flow in $\text{Ni}_3(\text{Al}, \text{Hf})$ single crystals [13 - 15]. Since the newly proposed technique also allows for the measurement of the strain-hardening rate ($K = d\tau_i/d\gamma_p$, τ_i is the internal stress) during a relaxation test, it is believed to be a supplementary tool that will provide additional information on the role of strain-hardening on the flow stress.

This paper reports the application of this method on $\text{Ni}_3(\text{Al}, \text{Hf})$ single crystals. Experiments have been conducted at various temperatures and for different crystallographic orientations. In a first attempt, the strain-hardening rates (K) are compared with those directly deduced from the stress-strain curves (θ).

2. Experimental Procedures

Specimens and deformation conditions

The compression specimens were cut by electro-erosion from a single crystalline rod of $\text{Ni}_{74.8}\text{Al}_{21.9}\text{Hf}_{3.3}$ supplied by Professor D. P. Pope of the University of Pennsylvania. Prior deformation, the samples were polished on successively finer grades of diamond paste down to $0.25\text{ }\mu\text{m}$ and finally electro-chemically polished to obtain suitable slip trace analyses. The final sizes of the specimens were $2.9\text{mm} \times 2.9\text{mm} \times 6.5\text{mm}$. Four different compression axes (figure 1) have been tested. In addition, samples with orientation 4 have been deformed at various temperatures (84K, 293K, 423K, 573K). All experiments were conducted in a Schenck RMC 100 testing machine at constant crosshead velocity. The nominal strain-rate was $5 \times 10^{-5}\text{ s}^{-1}$. The deformation were performed up to a plastic deformation (γ_p) of about 10%. A detailed description of the deformation conditions has been already reported elsewhere [16].

Stress-strain curve and repeated stress relaxation

The resolved shear stress (τ_a) and plastic shear strain (γ_p) were calculated with respect to the primary octahedral slip system (111)[101]. The Schmid factors are given in Table 1. The strain hardening rate $\theta_p = d\tau_a/d\gamma_p$ were calculated by numerically differentiating τ_a with respect to γ_p .

Orientation	1	2	3	4
Schmid Factor	0.44	0.47	0.50	0.46

Table 1. Schmid's factor for octahedral (111)[101] slip.

Figure1 . Sample orientation in the standard [001]-[011]-[111] unit triangle.

The repeated stress relaxation technique consists of performing a series of stress relaxations of constant duration (Δt) that all start at the same stress level (τ_a). A single relaxation test is performed in stopping the crosshead of the deformation machine at a given stage of the flow stress. Then, the load on the sample is allowed to relax and is recorded as a function of time. If dislocation movement is thermally activated, it can be shown in the framework of the thermal activation theory that the variation of the stress as a function of time during the relaxation is given by [17]:

$$\Delta\tau_a = -\frac{kT}{V_a} \log\left(\frac{t}{c} + 1\right) \quad (1) \quad \text{with} \quad \Delta\tau_a < 0$$

where c is an integration constant and V_a is the sum of two terms, the effective activation volume (V_{eff}) and a strain-hardening term (V_h) defined as:

$$V_{\text{eff}} = -\frac{d\Delta G(\tau^*)}{d\tau^*} \quad (2) \quad \text{and} \quad V_h = V_{\text{eff}} \frac{K}{M} \quad (3)$$

In expressions (2), $\Delta G(\tau^*)$ is the variation of the Gibbs free energy during the thermal activation event and $\tau^* = \tau_a - \tau_\mu$ is the effective stress. In expression (3), $K = d\tau_\mu/d\gamma_p$ is the strain-hardening rate that accounts for the variation of the internal stress associated with the plastic deformation taking place during the relaxation and M is the elastic modulus of the specimen-machine assembly.

The correction term (V_h) can be determined from the relaxed stress drop at the end of each relaxation of a series by using the relation [13]:

$$(n-1) V_h = \frac{kT}{\Delta \tau} \log \left(\frac{\exp(-\frac{V_a \Delta \tau_n}{kT}) - 1}{\exp(-\frac{V_a \Delta \tau_1}{kT}) - 1} \right) \quad (4) \quad \text{with} \quad \overline{\Delta \tau} = \frac{1}{n-1} \sum_j^{n-1} \Delta \tau_j$$

where n refers to the number of the successive relaxation. V_a can be obtained from the first relaxation test and V_h can be deduced from the slope of a linear fit of the right hand side of eqn. (4) versus $(n-1)$. Therefore, the strain-hardening coefficient K is obtained by the use of eqn. (3) since V_h , V_{eff} and M are experimentally known.

3. Experimental Results and Discussion

Before proceeding with the strain-hardening coefficient K , it must be emphasised that all of the relaxation data were nicely fitted by the use of eqns. (1) and (4). During the relaxation and for the selected duration Δt of 30s [13], the stress (τ_a) was observed to decrease as a logarithmic function of time (eqn. (1)). The fits of the stress drops measured at the end of each relaxation of the series were linear (eqn. (4)). More details on these experimental results can be found in references [13 - 15].

For all deformation conditions, the strain-hardening coefficients (K and θ_p) exhibit a similar behaviour with increasing plastic strain. Both start high and decrease rapidly with strain up to a value of $\gamma_p \simeq 2\%$, where they become practically constant. But, K is always larger than θ_p . In the following, for the sake of clarity, we shall report K and θ_p values measured in the deformation domain where they are almost constant (*i.e.* above 2%). Note that in this deformation stage, the plastic strain-rate ($\dot{\gamma}_p$) can be as well considered as constant.

At room temperature, the ratio K/θ_p is of the order of 4 for all of the four orientations. It must be emphasised that θ_p , and consequently K , exhibit large values as compared to pure fcc metals deformed in single slip ($\mu/\theta_p \sim 2000$, μ is the shear modulus) or even in multiple glide ($\mu/\theta_p \sim 300$). For comparison, Table 2 reports the reduced values μ/θ_p that were measured for the four orientations at room temperature. Note that for orientations 1, 2, and 3 slip traces that belong to the secondary slip system have been observed, but not for the samples with orientation 4.

Orientation	1	2	3	4
μ/θ_p	50	140	95	65

Table 2. μ/θ_p ratio at room temperature as a function of the deformation axis.

Figure 2. Strain-hardening rate θ_p (o) and K/θ_p ratio (●) as a function of temperature (orientation 4).

For orientation 4, which has only been investigated at different temperatures, both θ_p and K increase with temperature but K increases more rapidly than θ_p . Figure 2 shows the strain-hardening rate θ_p and the ratio K/θ_p as a function of the temperature. Note that at 573K serrated flow was observed above $\gamma_p \approx 2\%$. This indicates an unstable dislocation substructure and therefore the stress relaxation tests have been performed.

The plastic deformation that takes place during a stress relaxation test is small (typically $\Delta\gamma_p < 10^{-3}$), so that the dislocation substructure can be considered as constant. Consequently, the high values of the strain hardening rate K suggest that strain-hardening during the relaxation is an inherent part of dislocation motion. In other words, strain-hardening is a direct consequence of the cross-slip mechanism and does not result from the dislocation interactions with substructure defects, such as APB tubes [18] or "debris" [7]. The fact that K is found larger than θ_p indicates that dislocations are "immobilised" much more rapidly under stress relaxation conditions (decreasing τ_a) than under constant strain-rate deformation conditions (increasing τ_a). Support for this description is found in the experiments of Thornton *et al.* [7] showing that, for a given range of temperature in the anomalous domain, a strain-rate decrease was initially accompanied by a decrease of τ_a followed by a transient domain of rehardening.

4. Conclusion

Preliminary results have been obtained on the strain-hardening that occurs during stress relaxation tests performed on Ni₃(Al, Hf) single crystals. They can be summarised as follows:

- 1) The strain hardening (K) that takes place under stress relaxation conditions is larger than the one produced under constant strain-rate conditions (θ_p).
- 2) At room temperature, the ratio K/θ_p is found to be approximately orientation independent ($K/\theta_p \sim 4$)
- 3) K/θ_p is found to increase with temperature, at least for orientation 4 which has been temperature investigated, only.
- 4) Strain hardening is interpreted as a direct consequence of the cross-slip mechanism.

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